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WELDING Al-Zn-Mg (7xxx SERIES) ALLOYS

- A LITERATURE REVIEW

A. Romeyn

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In this review an attempt is made to provide a basis for the understanding of the physical metallurgy and stress corrosion cracking behaviour of the alloys in the 7xxx series. In particular the interrelation between, microstructure, environment, and stress state is examined. Recent alloys have been improved by a better understanding of the effects of materials processing operations on the microstructure and residual stress state. Advances in welding techniques offer an opportunity to substantially reduce the residual stress level associated with welding thick sections. The technique of pulsed current metal inert gas (synergic) welding holds considerable promise.

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POSTAL ADDRESS: Director, Materials Research Laboratories
P.O. Box 50, Ascot Vale, Victoria 3032, Australia

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AUTHOR(S)	CORPORATE AUTHOR
ROMEYN, A.	Materials Research Laboratories P.O. Box 50, Ascot Vale, Victoria 3032

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The use of high strength aluminium alloys, especially the alloys of the 7xxx series, in the construction of light armoured vehicles, light weight bridging, and other equipment, is of interest to the designers of military equipment. However, in the use of alloys which exhibit the highest mechanical properties and provide superior ballistic protection a major problem has been exposed. Stress corrosion cracking has been found to occur in both the parent plate and the welded joint.

In this review an attempt is made to provide a basis for the understanding of the physical metallurgy and stress corrosion cracking behaviour of the alloys in the 7xxx series. In particular the interrelation between, microstructure, environment, and stress state is examined. Recent alloys have been improved by a better understanding of the effects of materials processing operations on the microstructure and residual stress state. Advances in welding techniques offer an opportunity to substantially reduce the residual stress level associated with welding thick sections. The technique of pulsed current metal inert gas (synergic) welding holds considerable promise.

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WELDING Al-Zn-Mg (7xxx SERIES) ALLOYS

- A LITERATURE REVIEW

1. INTRODUCTION

Aluminium alloys in which zinc and magnesium are the main alloying additions (7xxx series) are conspicuous amongst the commercial aluminium alloys because of their ability to be age hardened to strengths well in excess of the strengths that can be produced in other aluminium alloys (Table 1).

Two major classes of the 7xxx alloys have been in use for some time. The high mechanical properties have been exploited in the aircraft industry since the Second World War. The range of high strength aircraft alloys normally contains major alloying additions of 4.3-6.8% zinc, 2.5-3.3% magnesium, 0.5-2.0% copper, and minor additions (less than 1%), of chromium, manganese, and zirconium. The weldable alloys have a reduced alloy content; zinc is restricted to 4.0-5.0%, magnesium to 1.4-2.0%, while copper is limited to 0.2%. Minor additions of chromium, manganese, or zirconium may still be made. The lower alloying content of the weldable alloys results in a reduction in the maximum strength that can be achieved, consequently these alloys are usually referred to as the medium strength 7xxx alloys.

In recent years there has been an increasing military interest in reducing the weight of equipment. Light weight engineering equipment can offer useful economies in transport, and manpower, while light weight armoured fighting vehicles can offer increased mobility.

A comparison of the two medium-to-high strength materials that are the main contenders for lightweight equipment, various aluminium alloys, and various grades of steel, reveals that the aluminium alloys have an advantage over the high strength steels in that for equal strength a member is thicker and therefore stiffer and more robust. The use of thicker aluminium plates results in a considerable increase in rigidity [1]. As a consequence a number of purely structural stiffeners can be eliminated from designed

structures which can lead to additional savings in weight and production man hours.

7xxx series alloys can provide superior ballistic protection under all modes of attack when compared to other aluminium alloys, and can be superior to steel armour on a weight for weight basis especially against attack by high explosive shell fragments [2], (Fig. 1). The use of thicker plates and the relative ease of machining of aluminium alloys allows the use of rabetted or stepped joints which in turn allow easier fit up and less welding.

However, the unique attraction of the weldable medium strength 7xxx alloys is that the weld deposit and the regions of the parent alloy affected by heat from the welding process may naturally recover to approximately 80% of the ultimate tensile strength of the parent alloy. In other aluminium alloys the heat produced during welding causes a marked reduction in the strength of the weld zone (Table 1).

Despite these attractive properties the weldable 7xxx alloys (Table 2) are not widely used for structural applications, probably because of the limited knowledge concerning the stress corrosion cracking susceptibility of these alloys. This problem has been hard to resolve because stress corrosion cracking involves a combination of microstructure, sustained tensile stress and environment. Microstructures and residual tensile stresses are strongly influenced by the semi-fabrication process and heat treatments, welding further modifies the microstructure and stress state.

It is the purpose of this review to examine the developments that have recently been made in the understanding of the microstructures that are formed and the stress states that are developed during the processing of 7xxx alloys from, as cast billets to welded structures, and how these factors in conjunction with the service environment affect the final properties of welded structures, in particular the resistance to stress corrosion cracking.

2. MICROSTRUCTURE PROPERTY RELATIONSHIPS

2.1 Strengthening Mechanisms

The physical metallurgy of the alloys in the 7xxx series has been extensively discussed and reviewed [3] since the initial work of Eger [4], Sander and Meissener [5]. One essential attribute of a precipitation hardening alloy system is a temperature dependent equilibrium solid solubility characterised by increasing solubility with increasing temperatures. At temperatures above the solvus the alloying elements are completely taken into solid solution. When the temperature is lowered below the solvus, the solid solution becomes supersaturated and a second phase forms by solid state precipitation. The nature and dispersion of these precipitates determines the extent of hardening.

The various precipitation reactions which may occur during the decomposition of the supersaturated solid solution have been investigated by Ryum [6]:

(i) solid solution → η ($MgZn_2$) precipitates

The equilibrium phase in alloys with compositions in the range 4-6% Zn and 1-3% Mg is η($MgZn_2$). The equilibrium phase at temperatures above approximately 200°C is T($Al,Zn_{48}Mg_{32}$). The activation energy for this reaction is high, with the result that homogeneous nucleation with low precipitate density only occurs after long ageing times, e.g. 12 days, at 150°C. However, extensive precipitation can occur heterogeneously on grain boundaries and small intermetallic particles at temperatures above 145°C.

(ii) solid solution → "nuclei" → η'

The "nuclei" involved in this reaction were considered to involve vacancy clusters. This reaction was found to be predominant in material quenched to room temperature and aged for times as short as 5 sec. and then aged further at approximately 150°C. The "nuclei" formed on quenching to room temperature were reported to be independent of Guinier Preston (G.P.) zone nucleation and growth. The η' precipitates can nucleate homogeneously after an incubation period if the material is quenched directly to a temperature between 125°C and 145°C.

(iii) solid solution → G.P. zones → η' → η

This reaction describes the complete decomposition of the supersaturated solid solution. The formation of G.P. zones dominates at temperatures from 20°C to 125°C. Prolonged ageing leads to the successive transformation of the G.P. zones to η' precipitates and then to η precipitates.

Peak strength is achieved by an optimum distribution of both G.P. zones and η' precipitates. The incoherent, stable, η precipitate makes a negligible contribution to the hardening of these alloys [7].

2.2 The Effects of Materials Processing Operations

In commercial alloys many processing and production variables can exert a strong influence on the ageing process, and other microstructural features can be developed which may influence properties such as toughness and susceptibility to stress corrosion cracking. Recently the improvements which have been made in the performance of 7xxx alloys have been the result of an improved understanding of the complex and sometimes subtle, interrelations between materials processing operations and microstructure.

Insoluble compounds based on iron, silicon and other impurity elements may form during casting. In a uniaxially worked wrought product

these second phase particles break up during working and appear in the final product aligned along grain boundaries as "stringers". The presence of aligned particles can greatly reduce toughness in the short transverse direction [8]. More homogeneous properties can be obtained by preventing the formation of these stringers. A working operation, such as forging, prior to rolling can be used to break up insoluble particles and distribute them in planes normal to the eventual rolling direction. However, such operations add considerably to the cost of the product.

Gains in toughness may also be obtained by reducing the level of iron, silicon and other impurity elements, however, the limits to such reductions are set by cost and availability of high purity materials [9] and the effect that these elements may have on the precipitation kinetics of the alloy [10].

Precipitation of η phase ($MgZn_2$) during cooling from solution treatment may also result in low strength if a large number of coarse precipitate particles are formed. Quenching procedures are important in avoiding this effect, the solid solution formed during solution treatment must be quenched rapidly enough and without interruption to produce a supersaturated solid solution from which G.P. zones and η' precipitates can form. Transfer times from furnaces to quenching stations and the quenching rate must be sufficiently high to avoid slow cooling or reheating in the temperature range 400°C to 290°C where rapid precipitation can occur [11].

The tendency to form η precipitates during cooling from the solution treating temperature is commonly referred to as quench sensitivity. Additions of chromium and manganese, usually made to prevent grain growth and recrystallisation, increase the quench sensitivity because the intermetallic precipitates they form provide heterogeneous nucleation sites for η phase. Ryum [6] proposed that only intermetallic precipitates with incoherent interfaces were favourable nucleation sites. The zirconium intermetallic compound Al_3Zr was found to be completely coherent with the matrix [12] and not an effective nucleation site. Therefore, zirconium additions are to be favoured if grain growth and recrystallisation control are required in addition to a tolerance for slow quenching rates.

The size and distribution of intermetallics formed during homogenisation or other high temperature treatments are largely controlled by temperature and time. If a coarse distribution of large intermetallic precipitates is formed the mechanical properties of the alloy can be adversely affected. For example [13], a fine distribution of the chromium intermetallic with a mean diameter of 450 Å produced by low temperature homogenisation results in only moderate recrystallisation, highly elongated grains, and a transgranular failure mode. On the other hand a chromium intermetallic particle size of 1200 Å results in almost complete recrystallisation, coarse and only slightly elongated grains, and an intergranular failure mode.

After solution treating and quenching the 7xxx alloys age harden naturally at room temperature. However, in contrast to the relatively stable condition reached in a few days by the 2xxx alloys, the 7xxx alloys are considerably less stable at room temperature and continue to exhibit significant changes in mechanical properties over a number of years. Therefore, the 7xxx alloys are normally used in an artificially aged condition, either the peak strength condition (T6), or the overaged condition (T7x) in which some strength is sacrificed for improvements in corrosion resistance and dimensional stability.

Commercial ageing heat treatments are often compromises between time and cost factors and the probability of obtaining the intended properties, with allowances made for variations in compositions within a specified range and temperature variations within the furnace and load. Lack of control during heat treatment can lead to unacceptable variations in properties.

Recommended treatments to produce T6 type tempers provide adequate tolerance for normal variations encountered with good operating practice. On the other hand T7x type tempers involve changes in strength that occur significantly more rapidly at the second stage of the T7x precipitation heat treatment. Consequently, control of both temperature and time to achieve the properties specified for these tempers is much more critical than the control required in producing the T6 temper. The rate of heating from the first to the second ageing step must also be considered because precipitation occurs during this period.

3. STRESS CORROSION CRACKING

Stress corrosion cracking is a problem which afflicts a wide variety of ferrous and non-ferrous alloys. Cracking occurs when a susceptible microstructure is exposed to a sustained tensile stress and particular environmental conditions. The prevention of cracking can be achieved by the removal or reduction of any of these factors.

Water or water vapour is the key environmental factor in the stress corrosion of 7xxx alloys, the presence of halide ions has a strong accelerating effect. Suitable tensile stresses can result directly from in service loadings, interference fit assembly, manufacturing processes, quenching during heat treatment, and welding. The determination of what constitutes a susceptible microstructure in the 7xxx alloys is not easy because of the complexity of the ageing reactions and, to date, the inability to adequately resolve all the microstructural constituents, even with transmission electron microscopy.

Stress corrosion failure in these alloys is characterised by intergranular cracking. The crack propagates along apparently weakened grain boundaries which lie in a plane normal to the tensile stress. The mechanism by which the grain boundaries are weakened is not clear, however it is

generally accepted that, atomic hydrogen, grain boundary segregation of solutes, grain boundary precipitates, and the electrochemical effects of segregates and precipitates may be involved. In a study of embrittlement of 7xxx alloys after exposure to water Hardie et al. [14] concluded that the embrittling effect of water depended on the strain rate and the electrochemical potential of the alloy in water. Anodic potentials resulted in embrittlement which was attributed to an anodic dissolution mechanism. Embrittlement involving the absorption of hydrogen was observed at cathodic potentials. At the free corrosion potential of the alloys examined the two processes appeared to overlap at strain rates of less than 10^{-6} sec^{-1} when ample time was allowed for absorption and dissolution reactions to take place.

The role of minor environmental changes, such as the presence of halide ions in increasing the onset of embrittlement was attributed to the breaking up of the surface oxide film and hence improving access of hydroxyl ions to clean surface [14]. It was found that chloride ions could break up the surface oxide film at 20°C. In the absence of halide ions it was suggested that temperatures above 40°C are necessary to produce the surface films which allow easy access of hydrogen to the metal substrate. This observation may have implications for the behaviour of 7xxx alloys in tropical environments.

In order to overcome stress corrosion cracking it would be desirable to produce a resistant microstructure which would allow the alloy to be used in various environments and in various stress states. The development of a suitable microstructure would allow a more confident and less restrained use of the 7xxx alloys.

One successful means of producing a resistant microstructure is by the overaging heat treatments which are designated T7x tempers. The stress corrosion cracking resistance is much greater than the peak aged T6 temper although, depending on the precise treatment, some strength may be sacrificed. The main and most obvious microstructural modification is the increased precipitation of η phase throughout the structure, while avoiding coarse precipitation of η at grain boundaries which degrades mechanical properties.

Narasimha Rao [15] studied the microstructural differences between T6 and T7x tempers in copper bearing 7xxx alloys using a scanning transmission electron microscope (STEM) with an energy dispersive X-ray analyser. In material tempered to the T6 condition silicon, copper and zinc segregates in the grain boundary region were identified. Grain boundary precipitates rich in zinc, magnesium, and copper were found to be lenticular in shape and closely spaced. In overaged specimens the segregation of copper, zinc, and silicon was found to be markedly reduced and the grain boundary precipitates assumed a more rounded shape and increased size. Overageing reduces the metallurgical notch effect of the grain boundary precipitates.

An alternative heat treatment which also achieves a controlled precipitation of η phase is commonly referred to as retrogression and re-ageing [16,17]. It is claimed that resistance to stress corrosion cracking

can be enhanced while retaining the mechanical properties of the alloy that are normally achieved in the T6 temper.

The presence of a suitable distribution of η precipitates may be effective in reducing the effect of hydrogen and in this way produces an increased resistance to stress corrosion cracking. Incoherent η precipitates may be effective traps for hydrogen which has found its way into the structure. Rajan et al. [17] considered that incoherent precipitates formed irreversible hydrogen traps. They proposed that the accumulation of dislocations at the particle/matrix interface provides the trapping sites for hydrogen. Coherent precipitates were not considered to be effective hydrogen traps.

The presence of a suitable distribution of η precipitates also effects the formation of the dislocation substructure during deformation. Incoherent precipitates, such as η phase, are not sheared by dislocations, therefore dislocations can only move by looping mechanisms with the result of producing a uniform distribution of dislocation tangles. Coherent precipitates, such as G.P. zones, can on the other hand be sheared by dislocations which in turn leads to the formation of slip bands. These slip bands, planar dislocation arrays, have been identified [18] as paths along which large amounts of hydrogen can be transported.

The partially coherent η' precipitates formed during duplex ageing treatments also prevent the formation of slip bands. This behaviour may explain why microstructures containing substantial numbers of η' precipitates show an improved resistance to stress corrosion cracking [19]. However, because the partially coherent precipitates are not as effective as incoherent precipitates in trapping hydrogen the improvement is not as great as that observed in overaged materials.

In summary it appears that environmentally induced intergranular failure can occur in 7xxx alloys providing the grain boundaries have a significant tensile stress applied normal to the plane of the boundary, and a critical concentration of hydrogen in the grain boundary region is exceeded [18]. Corrosion fissures resulting from the dissolution of lenticular grain boundary precipitates may increase the tendency to embrittlement by enhancing the stresses normal to the grain boundary below the notch root.

Materials processing operations must be aimed at the provision of hydrogen traps, the prevention of easy diffusion paths for hydrogen, and the prevention of anodic grain boundary regions and precipitates. Factors which have been found to influence the stress corrosion resistance of 7xxx alloy plate [19]; grain structure, quench rate from solution temperature, ageing temperature and time, composition, can be explained by their effect on the stress state normal to grain boundaries, on the size and distribution of all precipitates, and on the electrochemistry of the microstructural constituents.

4. WELDING

The procedures for the welding of aluminium alloys are well established. Defects such as lack of root, interrum, and sidewall fusion; porosity, oxide inclusions, and poor bead shapes can be easily controlled by the application of the correct welding procedure [20]. A special feature of metal inert gas (MIG) welding aluminium-magnesium and aluminium-zinc-magnesium alloys is that normal and long arc lengths (over 6.4 mm) will produce weld deposits which display fine surface porosity. This problem can be alleviated by reducing the arc length to about 4 mm and providing adequate gas shielding. A gas mixture of 75% helium 25% argon can be beneficial in the welding of 7xxx alloys because it improves the wetting characteristics of the material and alleviates undercut.

The MIG welding process is easily adapted to a mechanical operation where the gun is mechanically held and traversed along a joint at a constant speed. Mechanising the welding process has the advantages of better gas shielding coverage, more uniform weld deposits with fewer defects, more uniform penetration, higher welding speeds resulting in less distortion, and higher welding currents reducing the number of weld runs.

However, because the gun is mechanically held and traversed in a steady straight line, the advantages of the oscillation techniques practised during manual welding are absent. The situation may be remedied by the use of higher welding currents. Higher currents produce larger weld pools and the vigorous action of the more energetic arc agitates the enlarged pool sufficiently to prevent the formation of porosity. If the extra heat produced by the more energetic arc is likely to cause problems it may be necessary to use a mechanical oscillating head on the traversing unit.

The use of power sources which can generate a pulsed current waveform have become established in the MIG welding of aluminium alloys and they extend the previously limited scope of conventional, manual, and mechanised MIG processes, and bridge the transition zone between the tungsten inert gas (TIG) and MIG processes [21,22]. The pulsed MIG process offers a fast weld metal deposition rate with a substantially reduced overall heat input while maintaining a controlled (spray) mode of metal transfer. Large diameter wires, e.g. 1.6 mm, can be used with low average current values for the welding of a wide range of thicknesses, and the welding of joints between dissimilar thicknesses, e.g. 1.6 mm to 12.7 mm, while maintaining a uniform root penetration.

The heat generated during fusion welding has two very important effects on a welded joint; it can cause the formation of residual stresses in the region of the joint, and it can alter the microstructure of the parent material adjacent to the weld fusion zone. The strength of the heat affected zone in work hardened aluminium alloys, such as the 5xxx type, is reduced to that of the alloy in the soft annealed state. Overageing in responsible for appreciable reductions in the strength of the heat affected zones of precipitation hardened alloys such as the 6xxx type. The ability of the weld

zone in 7xxx type alloys to respond to natural ageing allows a considerable recovery of strength after welding, providing the composition of the fusion zone (determined by filler composition and dilution) is such that it will respond to natural ageing and the thermal history of the heat affected zone does not result in extensive overageing.

4.1 Structural changes in the heat affected zone

In alloys which contain and rely on various precipitates for hardening several reactions may occur in the heat affected zone [6].

(i) enhanced rate of growth of all precipitates
(ii) reversion of some precipitates
(iii) transformation of precipitates to more stable forms
(iv) the direct nucleation of more stable precipitates.

These reactions and the extent to which they occur are dependent on the structure of the alloy to be welded (chemical composition, state of precipitation, defect structure), and the thermal history (temperature, and time at temperature) which in turn is affected by the welding method and joint type.

The microstructural changes that can occur in heat affected zones in 7xxx alloys have been extensively investigated by Ryum [6]. A region adjacent to the fusion zone was identified where temperatures during welding were high enough for a sufficient period of time to dissolve all the precipitates present within grains (sometimes referred to as the white zone because of its etching behaviour [23]). The η' and η precipitates within grains were rapidly dissolved (less than 10 sec) at temperatures between 300 and 350°C, η precipitates on grain boundaries and sub grain boundaries dissolved when the temperature exceeded 400°C, at 350°C they remained stable and coarsened rapidly. On subsequent cooling various precipitates may form as the supersaturation increases, the formation of suitable hardening precipitates is dependent on the quench sensitivity of the region. It may be possible that grain boundary segregation following dissolution reactions may alter the quench sensitivity of grain boundaries leading to the enhanced precipitation of η phase at grain boundaries.

An outer region was also identified where the time-temperature conditions are such that some types of precipitates grow or transform to more stable forms, while others dissolve. At temperatures between 200 and 300°C η precipitates in the matrix and on heterogeneities coarsen very rapidly and η' precipitates dissolve within 10 to 30 secs at temperature.

In the remainder of the heat affected zone where the temperature does not exceed 200°C various dissolution and precipitation reactions can take place. Some of these reactions are rather sluggish so time and temperature and the state of precipitation before welding are important. Under slow heating rates precipitates would be expected to grow and transform to more stable forms, i.e. G.P. zones $\rightarrow \eta' \rightarrow \eta$. However, if a structure containing G.P. zones is heated rapidly the G.P. zones may become unstable and dissolve

even though this may increase supersaturation. This process, known as reversion, allows some supersaturation to be retained in the alloy after welding which can then respond to natural ageing and produce new G.P. zones. The recovery of strength (up to 80% of the ultimate tensile strength) accompanies the natural precipitation of these new G.P. zones

4.2 Welding problems peculiar to 7xxx alloys

Contributors to the "Proceedings of the Select Conference on Weldable Al-Zn-Mg Alloys" [24] identified four major potential problems which may arise during the welding of these alloys:

(i) solidification cracking
(ii) joint softening, not recoverable on post weld ageing
(iii) poor weld zone ductility
(iv) susceptibility of joint to stress corrosion cracking.

One of the major attributes of Al-Zn-Mg (7xxx) system which renders it particularly suitable for welding is the existence of various weld metal compositions that have similar ageing characteristics to the parent alloy while maintaining a high resistance to solidification cracking. The results of welding trials indicate that only when the magnesium content in the weld metal is lower than 2-2.5% will solidification cracking be a problem [25].

During the initial development of the weldable alloys it was anticipated that the mechanical properties of the weld zone would increase as the alloy content of the weld metal was increased, and the limit would be set by the onset of solidification cracking or the loss of notch toughness in the weld metal [26].

However, in practice, it was found that the loss of properties in a welded joint was associated with failure in the liquated region of the heat affected zone. When the parent alloy adjacent to the fusion zone experiences high heating rates the phenomenon of non equilibrium melting must be considered [27]. If grain boundary segregates and precipitates such as η phase ($MgZn_2$) are heated at a rate which does not allow sufficient time for diffusion to homogenise the structure, incipient melting may result in the liquation of grain boundaries. Microcracks may form in the liquated regions if hydrogen and/or sufficient strain is present. Also because of the change in composition in these regions toughness can be seriously impaired following ageing [26].

Control of liquation and liquation cracking can be achieved by control of grain size, residual impurities, degree of homogenisation, and alloy content. The copper content of the parent alloy has a significant effect on liquation cracking, which is why the copper content of weldable alloys is restricted to less than 0.2% rather than the normal addition of approximately 2% to the so called aircraft alloys.

The ageing response of the region of the heat affected zone which has been solution treated may be affected by the welding processes. A deterioration in ageing response has been noted by Rogerson [28] after multipass welding. However, the mechanism of this deterioration has not been clarified. It may be possible that the heating and cooling cycles experienced by the heat affected zone during multipass welding favours the growth of η precipitates at the expense of G.P. zones. The creation of solute segregation at grain boundaries in the heat affected zone during welding may also enhance the precipitation of η phase by increasing the quench sensitivity of the grain boundary region.

4.3 Stress Corrosion Cracking Associated with Welds

Welding may affect the stress corrosion resistance of fabricated structures in two ways. Firstly, the residual tensile stresses in the structure may be enhanced as a result of both the welding process and the stress concentration provided by the joint itself and its associated defects. Secondly, a susceptible microstructure may be produced in the fusion zone or the heat affected zone.

Normally, stress corrosion failures in the weld zones of 7xxx type alloys have been found to initiate and propagate in a region of the heat affected zone [19, 29], commonly referred to as the white zone. The microstructural features of the white zone are typical of a recrystallised structure. This recrystallised structure is caused by the large amounts of heat produced during some welding procedures.

The movement of high angle grain boundaries during recrystallisation can sweep out fine intermetallic compounds [29] necessary for the heterogeneous nucleation of η precipitates within grains and annihilate the vacancy nuclei necessary for the nucleation of η' precipitates [30]. The overall result of this movement is the formation of a very susceptible microstructure, comprising G.P. zones in the grain interiors with large closely spaced η precipitates at the grain boundaries. The movement of grain boundaries also means that some boundaries will become oriented normal to the residual tensile stresses that may form across the weld zone. These stresses arise from shrinkage during cooling after welding [31], and by the dimensional change (contraction) which accompanies the formation of G.P. zones from a solid solution during ageing [32].

4.4 Techniques for Reducing Stress Corrosion Cracking Associated with Welds

Pulsed MIG welding offers an opportunity to substantially reduce the overall heat input, and therefore the residual stress levels, associated with controlled metal transfer and the production of sound welds. Pulsed current MIG welding techniques are however, not favoured because of the complexities associated with adjusting the power supply when even minor changes are made to the welding procedure.

Research at the Welding Institute [33] has determined that a uniquely stable operating condition is obtained if the four basic current waveform parameters (pulse amplitude, pulse duration, background amplitude and background duration) are all directly related to the wire feed speed. The control of the pulse parameters as a function of wire feed speed can be accomplished with a microprocessor controlled transistor power supply and a digital wire feed unit. This technique, termed synergic pulse MIG welding, greatly simplifies setting up given working conditions since once the electronic controls are preset (and these remain constant for a given wire and shielding gas) the mean current level can be simply and continuously adjusted by the wire feed speed.

With the synergic pulsed welding technique it is possible to correlate transverse gun oscillation with that of the pulse controlled metal transfer. Metal transfer and arc heating may then be programmed to occur at the most advantageous positions with respect to the joint type and geometry.

Where stressing in the short transverse direction (through thickness) of thick plates is a problem, joint design, fit up, and the restraint employed during fabrication, should be examined and if possible altered to reduce stress levels. It may be pertinent to examine the recommendations documented for the case of lamellar tearing in thick steel plates [34] which also deal with the problem of stressing in the short transverse direction.

Attempts to improve the resistance of the weld zone to stress corrosion cracking by the addition of minor alloying elements has met with limited success [29]. The addition of approximately 1% silver to the filler was reported to markedly increase the resistance of the weld zone (including the white zone) only if the welded section was solution treated and artificially aged after welding.

5. SUMMARY

Initial experience with several 7xxx alloys (Table 2) in the demanding applications of light weight military bridging and armoured vehicles exposed the problem of stress corrosion cracking. An analysis of failures which occurred in these structures found that excessive levels of residual tensile stress produced during plate manufacture, assembly, and welding, was the factor most responsible for the high degree of susceptibility to stress corrosion cracking.

Since these first generation structures were manufactured, further advances have been made and more are possible, viz, in alloy design, materials processing, structural design, and welding processes, which may alleviate the problem of stress corrosion cracking.

The advances that have been made in alloy design and materials processing include, reductions in quench sensitivity through the control of alloying additions, avoidance of recrystallisation following hot working processes, stress relief after quenching, and heat treatments such as duplex ageing and overageing designed to produce a microstructure less susceptible to cracking. Also an appreciation has been gained into the effect of the compromises inherent in manufacturing which relate to, costs, the availability of suitable plant, and experience in producing limited tonnages of special alloys.

Welding and structural design is an area in which further advances are likely. It would be desirable to obtain an understanding of the nature of residual stress in a structure as a function of joint design and welding process. Welding processes such as synergic pulsed MIG with synchronised transverse oscillation offer the possibility of very high quality, high productivity, welding at markedly reduced heat inputs, and therefore with lower accompanying residual stresses. These processes may also lead to improved 'out of position' mechanised welding procedures which, in the future, may be capitalised on by robotic work stations to allow design freedom from straight down hand runs.

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TABLE 1

TYPICAL MECHANICAL PROPERTIES OF ALUMINIUM ALLOYS,

PLATE AND WELDED JOINTS

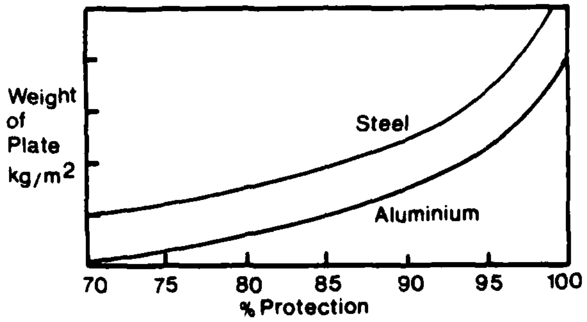
(From Alcan Plate Ltd and Aluminium Federation data sheets)

ALLOY DESIGNATION	TEMPER	PLATE PROPERTIES			BUTT WELDED JOINT		
		0.2% Proof MPa	UTS MPa	Elong. %	0.2% Proof MPa	UTS MPa	Elong. %
1200 (99% Al)	0	-	87	30	42	83	20
	H4	-	125	6			
	H8	-	140	-			
3103 (Al 1 Mn)	0	-	110	25	77	122	16
	H4	-	157	7			
	H8	-	175	-			
5083 (Al 4.5 Mg 0.75 Mn)	0	125	312	16	127	288	11
	H4	270	375	6			
6082 (Al 1 Mg 1 Si)	0	-	135	-	115	185	8
	T4	115	200	15			
	T6	240	295	8			
					Note: Properties for 7xxx alloys are for specimens nat. aged 30d. after welding.		
7020 Al 4.5 Zn 1.2 Mg	T6	360	420	13	220	320	9
7019 Al 4 Zn 2 Mg	T6	360	400	14	220	320	8
7017 Al 4.5 Zn 2.5 Mg	T6	435	485	11	220	340	8

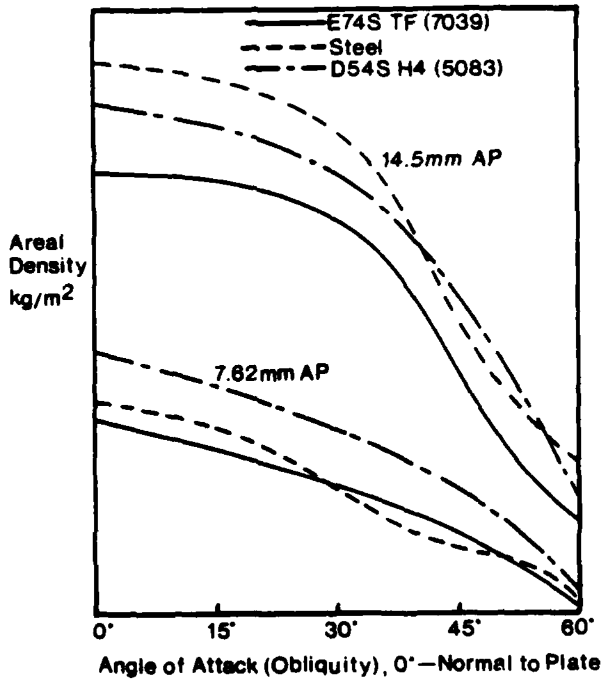
TABLE 2

7xxx ALLOYS OF DEFENCE SIGNIFICANCE

ALLOY DESIGNATION	NOMINAL COMPOSITION									USES
	Zn	Mg	Mn	Cr	Cu	Ti	Zr	Fe	Si	
7039	4.0	2.8	0.25	0.2	0.1m	0.1m	-	0.4m	0.3m	Early generation armour currently superceded.
7019 DGFVE 232 Alcan G74S	4.0	2.0	0.35	-	-	<.05	0.15	<.3	<.2	Light weight bridging
7020 Alcan D745	4.5	1.2	0.3	0.13	-	-	-	0.2	0.2	Used in Europe as Armour, compromise between strength and freedom from problems such as S.C.C.
7017 Alcan E74S	4.5	2.5	0.25	0.25m	0.25m	0.15m	0.25m	0.4m	0.3m	Highest strength armour still suitable for welding.



(a) Protection against 155mm H.E. fragments at 27.5m



(b) Protection against armour piercing rounds

Figure 1 A Comparison between Steel and Aluminium Armour, after Budd(1).

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